EROSION AND WEAR OF CERAMICS

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1. INTRODUCTION

Ceramics are among the hardest materials known to man. They consequently have great potential in applications where high resistance to surface wear and erosion is of prime importance. This same class of materials shows a unique capacity to withstand extremes in service conditions, notably at elevated temperatures and in generally hostile chemical environments. Coupled with the fact that ceramics tend to derive from the Earth's more plentiful elements, these attractive properties would appear to provide an ideal basis for industrial development of high-performance materials. Several areas may be cited where ceramic components are being successfully employed: windows in optical and infrared transmission regions (e.g. lasers, radome nose cones); blades for gas turbine engines; nuclear waste containment; photovoltaic panels for solar energy systems; prosthetic implants; etc.

However, there is one overriding factor which limits the ultimate usefulness of ceramics - brittleness. This brittleness is most strongly reflected in the notorious weakness of ceramic components under tensile loading; typically, μm-scale "flaws" are sufficient to render these materials far too susceptible to catastrophic fracture for widespread adoption in structural engineering. Designing with ceramics accordingly becomes a question of containing the nucleation and growth of microcracks. Nowhere is this last point more evident than in the erosion and wear properties, where local chipping about particle-contact sites constitutes the dominant mode of surface removal [1]. A proper understanding of our topic accordingly requires us to investigate the micromechanics of fracture in highly concentrated contact stress fields [2].

In this paper the basic mechanisms of material removal in multi-particle contact situations are reviewed. This is done in the context of "indentation fracture" analysis. A central element in the description is the due recognition given to the essential role of precursor deformation as the underlying driving force for the chipping mode. In consequence of this emergence of a plasticity component in an ostensibly brittle process, the theory of erosion and wear incorporates *hardness* as well as *toughness* as characteristic material parameters: hardness quantifies resistance to deformation, toughness quantifies resistance to fracture. Routine indentation testing provides information on both these parameters, and thereby sets the foundation for predetermining the material response under prospective erosive conditions.

2. BASIC MECHANISMS OF SURFACE REMOVAL

Considerable effort has recently been devoted to the study of the damage patterns produced on brittle surfaces contacted with hard, sharp particles [3-8]. These studies include observations by conventional optical and scanning electron microscopy, and by the more powerful techniques of transmission electron microscopy in the work by Hockey and co-workers. Particular attention has been given to the cases of normal particle impact (erosive wear) and translational particle sliding (abrasive wear). A conclusion of major importance to be drawn from the above studies is that the basic features of the damage pattern remain remarkably insensitive to the conditions of contact. These features are shown schematically in Fig.1.

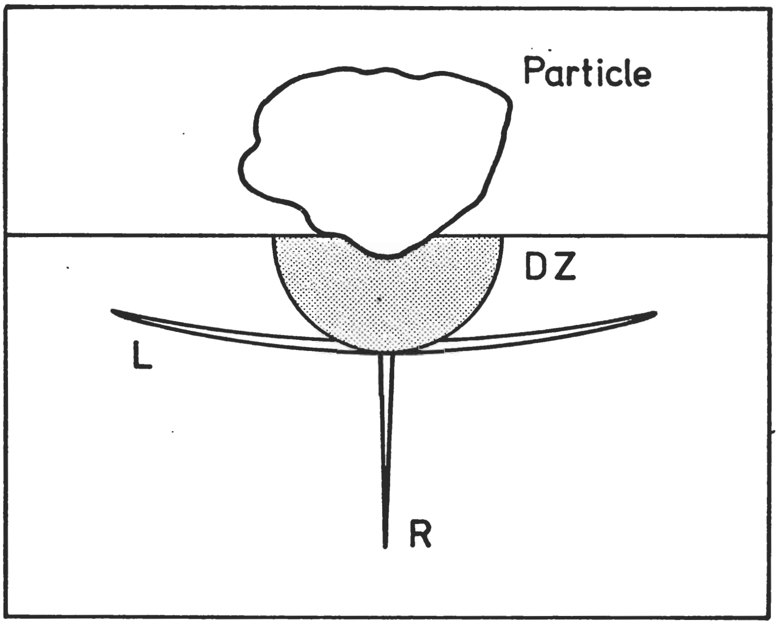

Fig. 1. Side view of contact damage, showing deformation zone (DZ) and attendant radial/median (R) and lateral (L) cracks.

Consider first the deformation component of the damage. This occurs in the volume of concentrated shear and hydrostatic compressive stresses immediately surrounding the contact area, particularly about penetrative corners or edges. The most obvious manifestation of the deformation process is the residual surface impression at the centre of the contact site. From the transmission electron microscopy studies the nature of the deformation is readily identified [3,4,8]: intense bands of dislocations and twins, highly localised in the central region, indicate a predominantly shear-operated plasticity mode. Associated with the intense, irreversible deformation is a strong residual stress field in the surrounding elastic material; it is this field which is the source of the ensuing crack driving force. A measure of the level of intensity of the deformation is given by the fact that in high-velocity contacts the near-surface layers (where the plastic work rate is greatest) tend to melt, even in materials such as alumina with melting points well in excess of 2000 K [9].

Next consider the fracture component. It is observed that there is generally a threshold in the loading, above which cracks begin to emanate from the deformation zone and grow along tensile stress trajectories into the surrounding elastic material. The cracks are of two basic types: (i) "radial/median" cracks, which tend to form on median planes containing the load axis and to leave characteristic radial traces on the contacted surface [2,10]; (ii) "lateral" cracks, which spread sideways from near the base of the deformation zone to form a saucer-like fracture approximately parallel to the surface [2,11]. The radial/median is the more penetrative of the two crack systems, and is therefore pertinent to strength properties in that it represents a possible source of premature component failure. The lateral crack, on the other hand, represents the more favourable configuration for chipping, and is consequently of greater interest here.

Broadly, the potential chip volume (depending on the incidence of neighbouring damage sites) may be reckoned as that volume encompassed between the lateral crack and the adjacent surface.

Two observations are of key importance in establishing the role of plasticity in the microfracture process. First, the damage configuration is remarkably independent of the initial state of the material surface. Cracks are produced even on surfaces of the highest perfection, indicating that the near-contact deformation must create its own flaw nuclei. (Traditional fracture theory supposes that flaw nuclei pre-exist in brittle solids.) Second, the crack growth which ultimately produces chipping can continue *after* the contact event, demonstrating that residual driving forces must be acting. Thus the deformation plays an active part in both *nucleating* and *propagating* the critical fractures in wear and erosion.

3. INDENTATION ANALYSIS OF DEFORMATION/FRACTURE DAMAGE

As mentioned above, the damage pattern in sharp-particle contact is not strongly sensitive to specific conditions of contact. This allows for the controlled study of the deformation/fracture micromechanics using standard indentation techniques. It is found that the Vickers diamond pyramid indenter used in routine hardness testing of metals is admirably suited to simulation of the "typical" contact event. In normal, quasistatic loading the indentation produces a well-defined, symmetrical pattern which is particularly amenable to measurement. Thus in Fig. 2 the scale of the deformation zone is characterised in terms of the impression half-diagonal

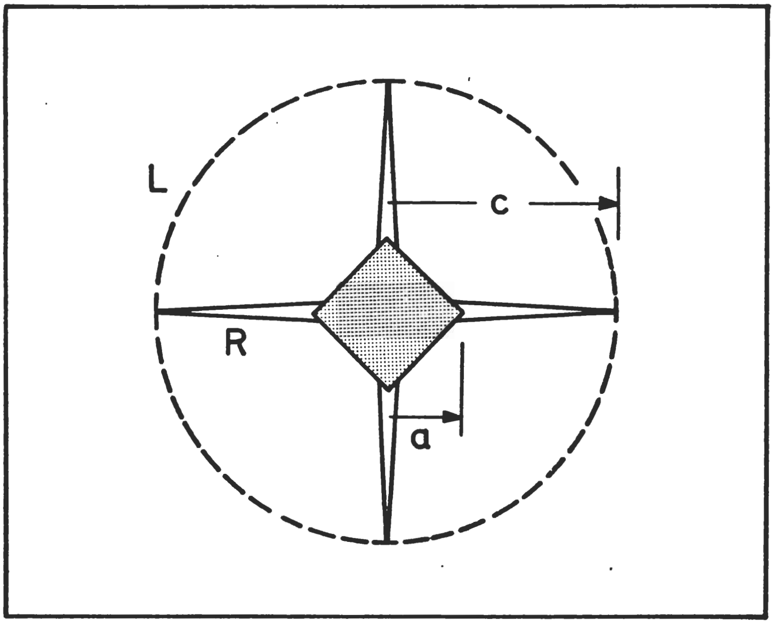

Fig. 2. Surface view of Vickers indentation pattern, showing characteristic deformation/fracture dimensions (a measured from diagonal of residual impression, c measured from surface trace of radial/median crack system).

a, and the scale of the cracking in terms of the radial dimension c. Note that both characteristic dimensions are obtainable from surface views alone, an obvious advantage in dealing with materials that are opaque.

Theoretical analysis of the indentation pattern in Fig. 2 gives rise to two simple relations for the scale of the deformation and fracture in terms of the peak contact load P [12,13]:

$$P/\alpha_0 a^2 = H \qquad \text{(1a)}$$

$$P/\beta_0 c^{3/2} = K_c \qquad \text{(1b)}$$

where H and K_c are constants which define material hardness and toughness respectively, and α_0 and β_0 are numerical factors. Physically, the hardness is a measure of the mean contact pressure, and, as such, relates directly to the characteristic stress level (e.g. "yield" stress) at which the subsurface deformation processes operate. Toughness as defined here corresponds to the intensity of the local field about the tip region of the equilibrium crack (critical stress intensity factor in fracture mechanics parlance).

It is interesting to note that although both a and c in Eq. (1) increase with P, as intuitively expected, they increase at different rates. This difference may be conveniently formalised by combining Eqs. (1a) and (1b) to obtain

$$c/a = (\alpha_0/\beta_0)(H/K_c)a^{1/2}, \qquad \text{(2)}$$

which shows that for any given material the crack size expands more rapidly than the impression size as the severity of contact, as determined by a, increases. Then we may define a "transition" contact dimension a^* corresponding to the condition $c/a = 1$, such that at $a < a^*$ the damage is effectively deformation-dominated (i.e. $c < a$), and conversely at $a > a^*$ the damage is fracture-dominated ($c > a$). From Eq.(2), the material dependence of this dimension is given by

$$a^* = [(\beta_0/\alpha_0)(K_c/H)]^2. \qquad \text{(3)}$$

Values for selected materials (including one non-ceramic) are listed in the Table below, using "calibrated" constants $\alpha_0 = 2$ and $\beta_0 = 7$ from Ref. 13.

The special significance of the dimension a^* is that it represents the scale of the damage event below which the material removal process is essentially ductile in nature, and above which it is essentially brittle. For instance, it is well known in the glass industry that if one requires a smooth, mirrorlike surface finish a fine-grained powder (sub-micrometre) must be used in the polishing operation. Conversely, if rapid removal is paramount, as in machining operations, relatively coarse abrading particles are necessary to activate the more effective chipping mode. (Note the relatively large value of a^* for steel, indicative of the almost exclusively ductile removal processes exhibited by most metallic and polymeric materials.)

Material	H (GPa)	K_c (MPa m$^{1/2}$)	a^* (μm)
Steel (medium strength)	5	50	1.2×10^3
Si_3N_4 (reaction bonded)	3.3	2.2	5.4
Si_3N_4 (hot pressed)	16	5.0	1.2
MgF_2 (hot pressed)	5.8	0.9	0.30
Glass (soda-lime)	6.2	0.7	0.16

In the context of optimisation of material properties for greatest resistance to erosion and wear, the best ceramics would appear to be those which are tough, yet soft. If one aims to design against the onset of the fracture mode it is the ratio K_c/H which needs to be maximised (to maximise a^* in Eq. 3). On the other hand, if it is argued that the prospective wear conditions are likely to be sufficiently severe that chipping is inevitable, the apparent requirement is for maximisation of K_c (to minimise c in Eq. 1b).

4. THEORIES OF EROSION AND WEAR

Given that the wear properties of ceramics are determined in the most unfavourable circumstances by an essentially fracture-controlled process, and that Fig. 1 reasonably represents that process, the indentation analysis above may be used to construct useful working theories for predictive purposes [1]. Basically, it is assumed that the volume of material removed per contact event is simply that which overlies the lateral crack; details as to how actual dislodgement of each fragment is effected (e.g. via interactions with neighbouring events) are not considered.

To illustrate the procedure, we take the case of sharp-particle erosion in normal impact. For the first step we note that the projected area of the lateral crack is ~ c^2, and that the depth of this crack is ~ a (recall that the crack initiation site is located at the base of the deformation zone), so that the volume removed per event is (neglecting constants of proportionality)

$$V \sim a\,c^2. \qquad (4)$$

Next, the impulsive load delivered in the event needs to be calculated in terms of some impact parameter for the particle, say the incident kinetic energy U_K. In the approximation that at maximum impression the bulk of this input energy has been absorbed as work of deformation in creating the plastic impression (the cracks thermselves actually absorb little energy in contact processes; as does the particle, provided it is harder than the target), the result

$$P \sim H^{1/3} U_K^{2/3} \tag{5}$$

is derived [14]. Combination of Eqs. (1), (4) and (5) then gives

$$V \sim (H^{1/9}/K_c^{4/3}) U_K^{11/9} . \tag{6}$$

The total volume removed is then found by summing Eq. (6) over all events.

This type of formulation, although by no means rigourous, provides a useful base for quantifying the important variables in the erosion problem, namely the material parameters of the target and the impact parameters of the projectile. Fig. 3 shows some data for three ceramics (see earlier Table) eroded with SiC particles, taken from unpublished data by M.E. Gulden and plotted in accordance with Eq. (6). The plotted points represent means and standard deviations for each material, covering a wide range of particle sizes (8-940 μm) and velocities (40-285 ms^{-1}). The theory appears to be capable of predicting the correct trends in material response. As with our conclusion in the previous section, maximisation of K_c offers the most effective route to optimal erosion resistance.

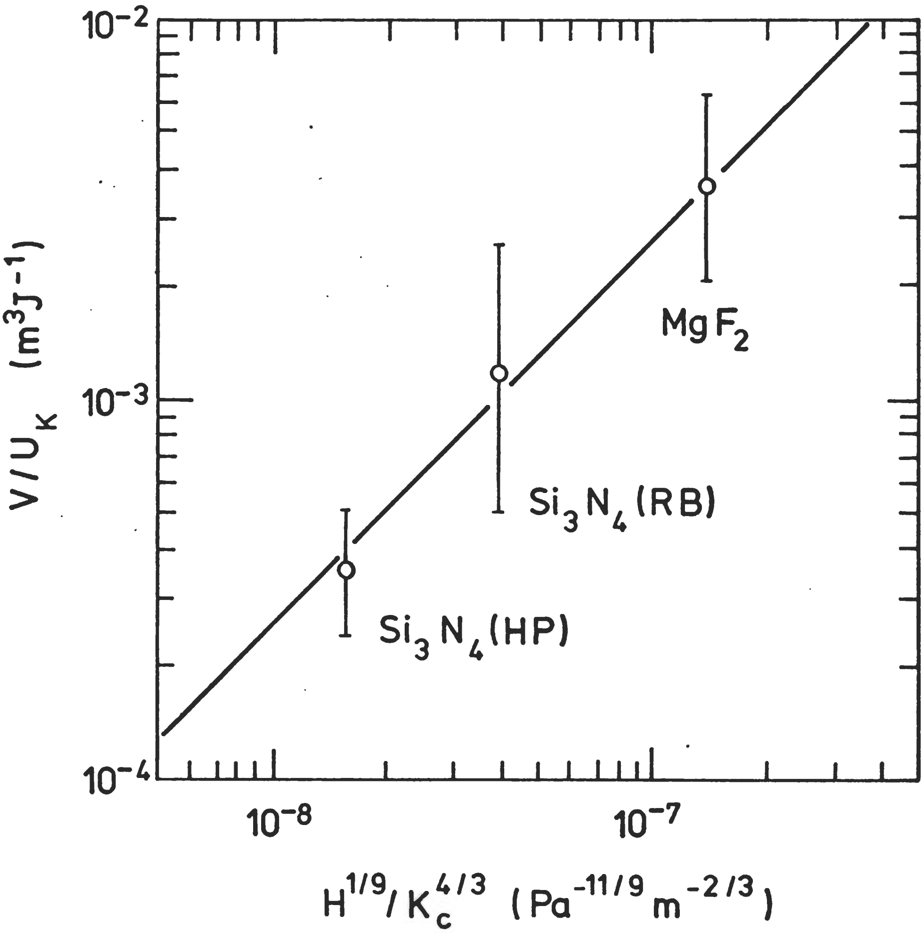

Fig. 3. Erosion data for ceramics impacted with SiC particles (courtesy M.E. Gulden)

5. CONCLUSIONS

Designing with ceramics against wear and erosion is essentially a matter of avoiding or minimising microfracture. The chipping mode of material removal is reasonably well defined, and thereby lends itself to analysis in terms of indentation theory. The theoretical analysis shows that the deformation/fracture response in sharp-particle contact can be characterised entirely by two basic parameters, hardness and toughness; and, moreover, that these parameters are readily obtainable from routine Vickers indentation tests. The primary requirement for high resistance to surface removal is large K_c, a secondary requirement is small H. Since K_c and H are such accessible, well-established parameters, the formulation described here is ideally suited to incorporation of extraneous variables (e.g. temperature and strain rate) via extensively documented data from materials evaluation programs.

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